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METALLURGICAL INVESTIGATION OF MAIN STEAM LINE WELDMENT CRACKING, LAKEVIEW TGS

Report No 89-233-K

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ABSTRACT

Failure of one weldment and cracking in three similar locations was due to "Type IV" creep damage in the heat-affected zone on the casting side of welds between cast and forged components. Primary cause of failure was excess "system" stressing, most probably due to improper functioning of the piping support system. Although ignored in piping design, the effects of load redistribution due to creep under service conditions should be compensated by periodic inspection and adjustment of the support system, as required. Because of systematic differences and some deficiencies in chemical composition, the castings could be expected to exhibit both inferior creep strength and creep ductility compared with the forgings. Based on metallographic examination and on published information, it is believed that the Type IV damage probably initiated below the OD surface. Developments in ultrasonic inspection techniques appear necessary in order to reliably detect the damage and to discriminate between real damage and innocuous weld metal defects.

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EXECUTIVE SUMMARY

METALLURGICAL INVESTIGATION OF MAIN STEAM LINE WELDMENT CRACKING, LAKEVIEW TGS

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Failure of a main steam line weldment in Lakeview Unit 3 and serious cracking in three similar locations was due to "Type IV" creep damage in the heat-affected zone on the casting side of welds between cast and forged components.

Based on detailed metallographic examination of the failed and damaged welds, and also on published information, it is believed that the damage probably initiated below the OD surface and would thus have escaped early detection by normal surface inspection techniques.

Primary cause of failure and cracking was excess "system" stressing, most probably due to improper functioning of the piping support system. Load redistribution due to creep in service, although ignored in piping design, needs to be compensated by periodic inspection and adjustment of the support system as-required.

Because of systematic differences, and also some deficiencies, in chemical composition, the castings would be expected to exhibit both inferior creep strength and creep ductility compared with the forgings.

Recommendations are made: (i) to supplement the present pipe weldment inspection program with some exploratory high temperature strain gauging of potentially susceptible terminal welds; (ii) to formulate a research program to examine the relationships between weld metal microstructure and in-service creep behaviour; (iii) to develop improved ultrasonic inspection techniques to reliably detect Type IV damage and to discriminate between real damage and innocuous weld metal defects.

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METALLURGICAL INVESTIGATION OF MAIN STEAM LINE WELDMENT CRACKING, LAKEVIEW TGS

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1.0 INTRODUCTION

In early March, 1989, a steam leakage in the vicinity of the West Main Steam Stop Valve caused a shut-down of Unit 3 at Lakeview TGS. Insulation removal and inspection revealed a large through-wall circumferential crack adjacent to the weld between a T piece and the main steam stop valve. Similar, but less advanced damage was then disclosed in the "sister" East Main Steam Line and also in both East and West lines of the identical Unit 4 at the same location, ie on the valve side of the weld between valve body and T piece. All these sections were removed and sent to Research for metallurgical investigation. Later, other "suspect" sections were removed from these units and also from Units 6 and 7.

In response to this serious failure, a program of steam line weldment inspections on all the Lakeview units was initiated and remains on-going, with Metallurgical Research being heavily involved in non-destructive metallographic assessment using the surface replication technique. The results of this major task, being co-ordinated by Dr. M.A. Clark, will be reported separately in due course.

This present report concerns the conventional "destructive" metallurgical investigation which has been carried out on the removed sections from failed, damaged and suspect locations in several units. The objectives of the work have been essentially two-fold, ie (i) to characterize the damage mechanism in detail, and (ii) to investigate the reasons for the damage.

1.1 Nature of the Failure

General details of the failure location are provided by Figures 1 and 2 which actually show the corresponding location and component details from Unit 4. As indicated, the main steam line dimensions were 13" (330 mm) OD and 2" (~51 mm) WT. All three components (pipe, T piece and valve) were specified to be 1.1/4 Cr - 1/2 Mo, with the T piece being a forging and the valve being cast. The T piece carried a by-pass line, unique to Units 3 and 4, which had not been used in practice. At the time of failure, operating hours on Unit 3 were ~ 88,000.

Figure 3 shows the crack in the weldment vicinity after removal of the T piece, the view being at the 6 o'clock (bottom) position which was approximately the mid-point of the crack. The region of through-wall penetration is shown in Figure 4. The extent of the crack was ~ 200° on the OD with the ID extent being estimated as ~ 180°.

The hanger shown in Figure 1 from Unit 4 was less robust in the Unit 3W location and had failed, the mechanism being readily identified as creep.

As noted, similar but less serious cracking was present in the Unit 3E and Unit 4E and W locations; in these cases, however, the adjacent pipe hanger supports were still intact.

Later inspection led to the removal of a further section of the failed line, at the weld between Pipe and Steam chest. Surface replication revealed cavitation damage in both the coarse and fine grain regions of the heat-affected zone (see later) and UT inspection suggested the presence of a major defect in this area. Subsequent sectioning and metallurgical examination, however, showed that the damage was essentially superficial with no significant through-wall penetration. In similar manner, a "spool piece" section of main steam line was removed from Unit 7 on the basis of UT indications but destructive examination only disclosed minor defects in the weld metal. No further description of work on these sections is felt necessary.

In Unit 6, no T piece was present but NDE indications of possible cracking at the weld between Main Steam Line and Stop Valve (casting) led to removal of this section. Again, metallographic examination showed no significant damage (see later) but some work on this weldment is included in this report in terms of affording a comparison between damaged and damage-free weldments.

2.0 LABORATORY INVESTIGATION

In line with the overall objectives noted earlier, the laboratory work fell into two main categories, ie (i) damage characterization and (ii) metallurgical investigation. The former has been based primarily on optical metallography on sections from failed or damaged weldments; the latter has involved chemical analyses, hardness testing, grain size measurement, scanning electron microscopy, and metal temperature estimation. In addition, a program of creep tests has been carried out on specimens from the Unit 6 weldment.

2.1 Damage Characterization

2.1.1 Failed Weldment (Unit 3 West)

As noted, the failure had apparently developed from the bottom (6 o'clock) position to grow ~ 200° around the wall. Longitudinal sections across the weld were made in the 6-9.30 sector for initial damage characterization.

Figure 5 shows the 6 o'clock section. The casting macrostructure is clearly evident, as is the multipass weld, the uniform structure of the forging and the two heat affected zones (HAZ). The crack has clearly followed the profile of the weld interface but is within the HAZ. Figure 6 shows the section at 9 o'clock where the crack is just through-wall. It is more evident here that the crack path is closer to the parent metal-HAZ interface than to the weld fusion line. A microsection taken ahead of the crack tip (Figure 7) reveals profuse cavitation damage in the "intercritical" zone of the HAZ, this being the transition between the fully-transformed, fine grain part of the HAZ and the partially-transformed parent material. The profuse cavitation confirms that the crack growth mechanism was creep-dominated and the path through the fine-grained region characterizes the damage as "Type IV" creep cracking. (This type of weldment creep damage is to be distinguished from the more common "Type III" cracking which occurs in the narrow coarse grain region of the HAZ adjacent to the fusion boundary).

Figure 8 shows the section at the ~9 - 30 position; this is near the crack extremity and, significantly, the crack here is sub-surface; this is confirmed by a microsection at the tip near the OD (Figure 9). Since crack extremities are usually indicative of the growth mechanism, this observation caused concern that the cracking may have initiated below the surface and thus would have been undetectable by surface replication or magnetic particle inspection. This aspect will be discussed later.

Details of the through-wall penetration were given by a further section taken at ~ the 8 - 15 position. Figure 10 taken at the ID shows that penetration resulted from linkage of the main crack from the OD with a small one initiating at the weld toe on the ID. Also, it is interesting that the main crack shows a "jog" near its extremity such that its path moves from the fine grain (Type IV) region to the coarse-grain (Type III) region. Cavitation at the tips of the Type III and Type IV parts of the crack is shown in Figures 11 and 12. Also near the jog, but somewhat further from the crack tip, Figure 13 shows profuse cavities in the fine grain zone while Figure 14 shows that some cavitation also extends into the adjacent parent material.

2.1.1 Damaged Weldments (Unit 3 East, Unit 4 East and West)

Sections through cracked regions of these welds revealed essentially similar, albeit less advanced damage characteristics to those just described for the failed weld, ie Type IV creep damage. Accordingly, it is felt appropriate to present only a few specific results of work on these welds.

Stages in the damage development are illustrated in Figure 15, a-d from the Unit 4 East weld; each micrograph is from a different section separated by ~ 1 cm. Figure 15a shows individual cavities in a zone of variable grain size, ie the fine grain/partially transformed intercritical part of the HAZ; 15b shows the beginnings of cavity linkage to form oriented microcracks, 15c shows a slightly higher degree of microcrack development and 15d shows the stage immediately preceding final linkage of microcracks to form a major defect.

It had become apparent that the early stages of Type IV damage involved profuse cavitation which, because of the fine grain size in the damage zone, might pass undetected except at magnifications approaching the limit of optical microscopy. Thus, Figure 16 from the Unit 4 West weld shows cavitation at 800X magnification which would probably not be detected at the more usual levels of inspection. At this stage, where cavities and particles are similar in size, it is not always easy to distinguish the two. Scanning electron microscopy is helpful in this regard and an SEM micrograph from the same weld (Figure 17) confirms the presence of cavities.

2.1.3 Defect Development

Based on the premise that the nature of the crack extremity could be used as an indicator of the initiation mode, particular emphasis was placed on mapping of the extremity using successive sectioning, not all on the same sample, in order to derive a possible description of defect development.

As shown in Figure 8 from the failed weld, the crack was clearly sub-surface near its extremity so efforts were made to trace the damage until its disappearance.

Figure 18 is from the Unit 3 East weld and shows that the crack front is non-uniform or wavy in nature. This region of the defect was located ~ 3 mm below the OD surface. Figure 19 shows what is believed to be the optically-detectable extremity of the same defect in the form of a short line of isolated cavities. The location of this area is indicated in Figure 20 ie, in the Type IV region ~ 11 mm below the OD surface.

By assembling the information from the various sections throughout the cracks, a possible defect development map was constructed and is shown in Figure 21. The map shows the postulated defect at successive stages, ie (i) origin, just below the OD surface; (ii) when first breaking the OD surface; (iii) when first penetrating the wall.

According to this figure, a fairly significant defect might exist before being detectable by surface inspection (replica, MPI). It must be emphasized, however, that this representation should be regarded as somewhat tentative.

2.1.4 Undamaged Weldments

As noted earlier, in-plant inspection of several welds using ultrasonics (UT) gave strong indications of possible cracking and led to the sections being cut out and sent to Research for metallographic examination. Prior to being sectioned, these samples were UT inspected under laboratory conditions and similar strong indications were obtained. Subsequent sectioning, however, revealed only minor defects in the weld metal with no evidence of Type IV creep damage. A section from Unit 6 typified this experience; the section through the region of maximum UT indication is shown in Figure 22. An optically visible defect was present in the weld, as indicated and shown in detail in Figure 23. Similar defects were also found to be giving rise to large UT indications in the spool piece removed from Unit 7.

The Unit 6 section was used for several purposes of comparison with damaged welds and was also used for the creep test program to be described later.

2.2 METALLURGICAL INVESTIGATION

Most of the metallurgical testing has been on the failed (Unit 3 West) weldment but, as noted, some comparisons have been made with the undamaged Unit 6 weldment and chemical analyses have been made on all the damaged and the Unit 6 weldments. For convenience, the results will be presented under two sub-headings, ie Failed Weld and Comparative Tests. The creep tests on the Unit 6 weldment will be described separately.

2.2.1 FAILED WELD

2.2.1.1 Metal Temperature Estimation

A relatively premature creep failure can be promoted by excessive metal temperature so it was important to examine this possibility. A fairly well established method is by means of steam side scale thickness measurement; several time-temperature-thickness relationships are available, notably those of Laborelec and Aptech; details have been given elsewhere/2/. For the 1.1/4 Cr-1/2 Mo, data are only available for the Laborelec method.

The measured scale thickness was 0.23 mm which, for the service time of 88,000 h, gave a metal temperature estimate of 538°C. Since this is the design steam temperature, overtemperature exposure did not appear to be a factor contributing to the failure.

2.2.1.2 Grain Size

Grain size was measured on the casting (valve body), forging (T-piece) and pipe sections of the failed assembly using the linear intercept method on photomicrographs. Mean grain diameters obtained were as follows:

Casting	Forging	Pipe
33.8 µm	30.1 µm	29.1 µm

Main interest is in comparison with the casting and forging; there is clearly a slight grain size difference and the possible sigificance of this will be discussed later.

2.2.1.3 Hardness

Hardness traverses were made at the 12 o'clock (undamaged) location using the Vickers hardness tester with 1 Kg load. Hardness was tracked from the weld metal, across the HAZ and into parent metal for the casting and forging sides of the failed weldment. Three traverses were made over a depth of ~ 10 mm just below the OD surface. Results are presented graphically in Figure 24.

In each case, the hardness falls significantly from the weld metal into the HAZ and there is also some hardness reduction with successive traverses away from the OD; this effect is slightly more pronounced on the casting side. Thus, in the region of failure (the Type IV region), the casting exhibited a slightly less uniform hardness over the sampled area.

For each weld, the maximum hardness difference between weld metal and parent metal expressed as the ratio minimum:maximum was ~0.6; the possible significance of this ratio will be discussed later.

2.2.1.4 Microstructure

Scanning electron microscopy was used to compare the microstructure across the casting and forging weldments in the areas corresponding to the hardness surveys. Micrographs were taken in weld, coarse grain HAZ, fine grain HAZ, partially-transformed and parent metal zones. Figures 25a and b and 26a and b show the fine grain and partially-transformed regions at 1000 X magnification for casting and forging respectively. No significant difference is apparent and the same was true for the other areas of the weldments.

2.2.2 COMPARATIVE TESTS

2.2.2.1 Hardness

Figure 27 shows hardness surveys across a damaged weldment (Unit 3 East - 4.30 position) and the undamaged Unit 6 weldment. Two surveys were made in each case, one close to the OD surface and one in the so-called "cusp" region below the final weld pass, about 10 mm below the surface. The Unit 6 hardnesses follow the trend previously shown for the failed (Unit 3 West) casting-forging weldment, ie maximum hardness in the weld metal with a signicant drop into the HAZ and a more gradual decline into the parent metal. In the Unit 3 East case, however, peak hardness on the casting side is in the HAZ; on the forging side, the near-surface hardness shows the same trend while the cusp region shows a slight decrease from weld to HAZ to parent

hardness. In terms of hardness differentials (see later discussion), the biggest differences were between parent and weld metal for the Unit 6 weldment while, in the Unit 3 East case, the biggest difference was between parent and HAZ in three of the four cases. For Unit 6, minimum ratios (minimum:maximum hardness) were in the range 0.63 - 0.70; for Unit 3 East, the range was 0.78 - 0.94.

2.2.2.2 Chemical Analyses

Table 1 presents chemical analyses for failed and damaged weldments (casting and forging side) and for the casting and pipe side of the Unit 6 weldment. Also given are the specified compositions believed to have been applicable at the time of Unit 3 construction. In addition to the specified elements, it was decided to include the trace elements, copper, arsenic and tin since these are potentially deleterious to creep ductility.

Scrutiny of these analyses reveals several trends, as follows:

i) Levels of carbon, manganese, phosphorus and sulphur are consistently higher in castings than forgings while the silicon level is consistently lower. These differences are in general accordance with slight variations in the casting and forging specifications for the 1.1/4 Cr - 1/2 Mo material.

ii) In all but one case, chromium levels in castings are higher than in forgings; some levels are slightly below the specified minimum.

iii) In all cases, molybdenum levels are lower in castings than forgings; in two cases, including the failed weldment, the levels are significantly below the specified minimum.

iv) In all the failed or cracked weldments, arsenic and tin levels, although acceptable, are 2 - 3 times higher in castings than forgings.

The significance of these results will be discussed later.

2.2.2.3 Creep Testing

In order to compare the creep properties of casting and forging/pipe weldments, a creep program was undertaken using the undamaged Unit 6 weldment (casting to pipe). This work is on-going and will be reported more fully in due course. It is, however, important to include the results to date in the present report since they have significant implications on the overall investigation.

Specimens were machined from each side of the weld as shown in Figure 28 so that the fusion boundary was normal to the tensile axis and also roughly in the centre of the gauge length. Isostress testing was carried out, initially at a stress of 50 MPa over the temperature range 657 - 630°C. This stress was believed to be service-realistic in terms of design and "system" loading (see discussion section). Later, in order to generate data at lower temperatures, a second series of tests was run at 100 MPa over the temperature range 620 - 580°C.

Results of these two test series are given graphically in Figure 29. Rupture lives ranged from 24 - 1486 h. All the data show a typical linear relationship between temperature and log time to rupture but an unusual difference is apparent at the two stress levels. Thus, at 50 MPa, the creep strength/resistance of the pipe side is somewhat greater than that of the casting side; at 100 MPa, however, the reverse trend is shown. These results, particularly those at 50 MPa, will be discussed later. It must be pointed out that none of these specimens failed by Type IV creep damage. In all but one case, rupture occurred in the parent material; one specimen failed in the HAZ but all ruptures were essentially ductile in nature.

3.0 DISCUSSION

3.1 THE TYPE IV PHENOMENON

There is no question that the failure in the Unit 3 East weldment and cracking in three "sister" locations was due to the so-called "Type IV" creep damage. Before the specific aspects of the discussing investigation, it is felt worthwhile to review briefly what is known about this form of damage.

Until very recently, there has been little awareness among North American utilities of the Type IV cracking phenomenon. In 1978, however, TVA experienced a failure in a 1/2 Cr-Mo-V pipe which was investigated in collaboration with Combustion Engineering and described in a 1982 publication/3/. Although the term "Type IV" was not used, the damage described was clearly of this type.

In Europe and the UK, however, the Type IV damage has been recognized for more than ten years and is believed to have caused the replacement of at least 40-50 joints in CEGB stations. The phenomenon has also been documented by operators in the USSR, the Netherlands and Germany. A detailed report of discussions held in April 1989 between Dr. S. Lawson of OHRD and representatives of CEGB and The Welding Institute (UK) was issued as a memorandum and is included in this report as Appendix A. This gives a comprehensive review of the CEGB/TWI understanding of the Type IV phenomenon together with some opinions expressed concerning the Lakeview failure. The reader is advised to read the appendix before proceeding further with this discussion.

3.2 PRESENT INVESTIGATION

3.2.1 Primary Cause of Failure

In general, premature creep failure is usually associated with excess metal temperature or excess stress. The steam side scale thickness indicated that design steam temperature had prevailed so it is pertinent to examine the stress situation.

Using the metal temperature 538°C, ISO creep rupture data for 1.1/4 Cr - 1/2 Mo/4/ can be used to estimate the stress to cause rupture in 88,000 h (the time of failure). However, based on CEGB predictions concerning Cr-Mo weldments (see Appendix) it appears appropriate to increase this time by 20% to 110,000 h. The predicted stres to rupture parent material in this time at 538°C is ~ 66 MPa.

For the given component geometry and steam conditions, the mean diameter hoop stress is 44.55 MPa. Since the failure was circumferential, axial stress is clearly more relevant. Nominal axial stress is half the hoop stress, although the design code implicitly permits the axial stress to equal the hoop stress/5/. This, plus consideration of possible extraneous loading, was the reason to select 50 MPa for the initial creep test series (see later).

From the above, it appears that axial (likely bending) stresses 3X the hoop stress had been present. Clearly, this is excessive "system" loading, strongly suggesting that the piping support system had not been functioning correctly.

It is not known whether the excess stress had been present throughout the service life or whether the 62 MPa is an "effective stress" due to stress build-up throughout life. If the latter were the case, clearly the stress in the late stages of life could have been considerably greater.

Recent CEGB studies have analysed in-service creep in piping systems and have emphasized the point that the effects of creep on the primary system loadings sustained by high temperature steam piping are ignored in design/6/. Because the loading does re-distribute under creep conditions in service, it is important that periodic inspection and adjustment of piping support systems be carried out to avoid stress build-up at susceptible locations, of which terminal welds are the prime example.

Incidentally, it felt that the failed support adjacent to the weld failure was more a consequence of general system deflection than a direct cause of the failure.

3.2.2 Damage Initiation and Development

One of the main concerns which arose during the investigation was the possibility that the Type IV damage can develop sub-surface and thus escape detection in the early stages. The concern was based on crack extremity observations and also on information in the literature. Thus, the crack extremity was clearly several mm below the OD surface. Although this is not proof of sub-surface defect nucleation, it is a strong indication since, once a crack is open to the external surface, it would appear unlikely that its leading edge would "tunnel" below this surface unless specific conditions existed which favoured this. As noted in the appendix (p 3), theoretical explanations of sub-surface growth have been offered. These explanations are based on the effects of differential creep rates in the different HAZ regions, and on the position of maximum hydrostatic (triaxial) stress. Also the weld profile, particularly regarding the "cusp" between the final pass and the previous passes influences the position of maximum creep damage; this has been discussed in the context of longitudinal seam weld cracking/7/. In the present investigation, the crack extremity and inferred origin were in this cusp region.

From the limited information in the literature it appears that, under pure axial loading, Type IV damage does tend to develop within the component wall/8,9,10/. The effect of bending stress, however, is to move the origin outwards so that, at some unspecified level of bending stress, the damage nucleates and develops from the surface. In the present case, it would appear that bending stresses were present but not at quite the level to cause surface nucleation. Although the derived "map" of defect development (Figure 21) suggested a substantial extent of sub-surface growth, it must be emphasized that this diagram is based on observations at the defect extremity ~ 90° from the origin; an implicit assumption of uniform stress around the pipe is made which, in the case of bending, may be inappropriate. Thus, if the bending stress were maximized at the defect origin, the damage might be OD-detectable earlier than as suggested by Figure 21. On the other hand, if the bending stress increased due to reducing load bearing area as the crack grew, the reverse might be the case.

3.2.3 Non-Destructive Inspection

The intensive program of weldment inspection following the failure has confirmed the ability of surface replicaton to detect cavitation damage in the early stages. The one proviso, noted earlier, is the need for high magnification to resolve cavities in the fine grain region of the HAZ. Magnetic particle inspection also has proved effective in detecting surface-breaking cracks.

The situation is less satisfactory, however, in regard to detection of sub-surface damage. The problem with UT appears to be in discriminating between real damage and innocuous weld metal defects. It appears that UT will detect an actual crack if present sub-surface. Such a crack was recently detected during the Unit 3 inspection program and confirmed by grinding; no surface damage was present. Most regrettably, it was not possible to remove the damaged area to permit laboratory characterization of this defect. On the other hand, several welds were removed on the basis of UT indications which proved misleading.

Work is in progress, both within OH and in other laboratories to develop UT procedures for this application so, hopefully, a more reliable technique may be forthcoming.

3.2.4 Metallurgical Investigation

A common factor in the failed and cracked weldments was that damage had developed on the casting side of the weld. Consequently, an objective of the metallurgical investigation was to examine whether cast material was inherently more susceptible to creep damage than wrought material (forging or pipe). As noted in the Appendix, outside opinion discounted this possibility, placing more emphasis on component design factors.

The creep resistance of low-alloy steels mainly derives from a dispersion of carbide particles in the microstructure and, to some extent, the strengthening can be related to hardness measured at ambient temperatures. On this basis, hardness variation is indicative of differential creep rate under service conditions. In weldments, therefore, softer regions of the HAZ are expected to concentrate creep strain.

As shown in Figure 24, significant softening occurred from the weld metal through the HAZ to the parent metal on both casting and forging sides of the failed weld. Since, in each case, the softest material was the parent metal, hardness alone does not determine creep damage location. It should be understood, however, that room temperature hardness is, to some extent, grain size-dependent and that the inverse dependency applies to creep resistance. Thus, grain boundaries strengthen (harden) the material at ambient temperature but effectively cause softening at creep temperatures due to their role in the creep deformation processes. Thus, despite higher measured hardness, the fine grain HAZ would be expected to be less creep-resistant than the parent material.

Recent work related to the longitudinal seam weld failure problem has shown a relationship between hardness differential and cavitation density as shown in Figure 30/7/. The hardness differential is the ratio of the minimum to maximum measured hardness in the different parts of the weld, whether between weld and HAZ, weld and parent, or HAZ and parent. Note that the cavity density increases markedly (log scale) as the differential increases over the range 0.9 - 0.8.

In the present investigation, the hardness differentials were determined in comparing damaged (Unit 3 East) and undamaged (Unit 6) weldments. However, the damaged weld values were significantly higher (in the range 0.8 - 0.95) than the undamaged values which were below 0.7. For the failed weld, low values (~ 0.6) were obtained on both casting and forging sides. Based on these findings, it would appear that other factors (such as stress differences) may be more important than hardness differentials in promoting creep damage.

The creep tests at the lower stress showed that the casting had somewhat lower strength than the pipe, though the effect was reversed at higher stress. In the former case, the stress was felt to be service-realistic (hoop stress + significant system stress) as discussed earlier, but the temperatures were high; in the latter case, the stress was excessive but the temperatures were lower, though still high relative to service conditions.

Concerning the 50 MPa casting results, extrapolation of the data down to 538°C (the service temperature) gives a predicted rupture life of ~160,000 h. At the time of weldment removal, service life of Unit 6 was ~75,000 h. Thus the total predicted life of the casting was 235,000 h. (Note, that since the creep specimens failed in parent material, it is appropriate to refer to casting rather than weld life).

Using ISO creep data for 1.1/4 Cr - 1/2 Mo/4/, the expected life at 538°C and 50 MPa can be determined; a value of ~229,000 h is obtained, ie virtually equal to the predicted life from the iso-stress tests. The implication is that the stress of 50 MPa was, in fact, very close to the "effective" service axial stress. As discussed earlier for the failed weld, this is substantially above the pressure-induced nominal axial stress, indicating the presence of significant "system" loading.

As noted, the mode of failure in the test specimens did not reproduce the Type IV cracking; it seems likely that some form of notched specimen geometry will be needed to promote this kind of failure in a uniaxially-loaded creep specimen. It is hoped to continue the creep work along these lines and report the findings when appropriate.

The most significant differences between castings and forgings emerged from the chemical analyses. These differences must be evaluated in terms of their potential influence on the creep properties, specifically the creep strength/resistance and the creep ductility. The strength relates to the rate of strain and the ductility detemines how much strain can be accommodated before the onset of cavitation.

In the Cr-Mo low-alloy steels, molybdenum is the main contributor to creep strengthening, by forming relatively stable carbide particles which impede dislocation motion. The analyses showed that, in all cases, the molybdenum level was significantly lower in the casting than in the forging or pipe; the biggest differences were in the cracked Unit 4 East weld and in the failed Unit 3 East weld; in these cases, the molybdenum levels were significantly below the specified minimum value. On this basis, there is evidence that the creep resistance of the castings was systematically lower than that of the forgings. This is supported by the evidence from the lower stress creep tests, although the Unit 6 weldment used for the creep tests showed only a minor difference in molybdenum levels between casting and forging; the implication is that the other cases where large molybdenum differences were present may have had correspondingly larger creep strength differentials.

Concerning creep ductility, this is inversely related to the propensity for intergranular cavitation which itself depends on the ease with which cavities can be nucleated on the grain boundaries. Nucleation is facilitated by reduction in the grain boundary surface energy and it is well recognized that this can result from segregation of various elements. Figure 31, taken from a review paper on intergranular failure in steel/11/ shows the roles of various elements in temper embrittlement, which results from exposure of previously tempered steels to temperatures in the range 375 - 560°C. This causes loss in room temperature toughness but is also responsible for promoting cavitation under creep conditions. Since the above temperature range encompasses the service conditions of power plant components, Figure 31 is equally applicable to creep embrittlement.

Referring to Figure 31, the following elements are identified as deleterious: silicon, sulphur phosphorus, arsenic, tin. With the exception of silicon, all these elements were systematically higher in the castings than in the forgings. Also, chromium is identified as a segregation enhancer while molybdenum is a segregation inhibitor. Since the castings had generally higher chromium and lower molybdenum, this is further evidence for reduced creep ductility. Although not a major factor, it is worth noting that the slight difference in grain size between casting and forging would slightly increase the grain boundary concentration of segregated elements (since the grain boundary area is inversely related to the grain size).

Several empirical quantitative relationships have been developed between chemical composition and propensity to temper embrittlement; one of these is the so-called "J-Factor", as follows:

$$J = (Si + Mn)(P + Sn) \times 10^4 \qquad (12)$$

Steels with J values ~ 100 are regarded as satisfactory; values ~ 300 indicate a strong tendency towards temper embrittlement.

Table II gives J values for the casting and forging side of the weldments examined. For each weldment, the casting J value is significantly higher than that of the forging, in three cases (including the failure) the difference being dramatic; all the casting J values are well into the embrittlement propensity range.

Thus, based on the chemical compositions, there is strong indication that both the creep strength and creep ductility of the castings were inferior to those of the forgings so that, other things being equal, creep failure would be predicted on the casting side of a casting-to-forging (or pipe) weldment.

4.0 CONCLUSIONS

1. Failure of one weldment and cracking in three similar locations was due to "Type IV" creep damage in the heat-affected zone on the casting side of welds between cast and forged components.

2. Based on metallographic examination and on published information, it is believed that the damage probably initiated below the OD surface.

3. Primary cause of failure and cracking was excess system stressing, most probably due to improper functioning of the piping support system.

4. Load redistribution due to creep in service is ignored in piping design but should be compensated by periodic inspection and adjustment of the support system, as required.

5. Because of systematic differences and some deficiencies in chemical composition, the castings would be expected to exhibit both inferior creep strength and creep ductility compared with the forgings.

6. Developments are needed in ultrasonic inspection techniques to reliably detect Type IV damage and to discriminate between real damage and innocuous weld metal defects.

5.0 RECOMMENDATIONS

1. Since Type IV damage is known to be promoted by excess system stresses, it is felt that some exploratory high temperature strain gauging of potentially susceptible terminal welds would be a useful supplement to the present inspection program involving replication and NDE.

2. The relation between weldment microstructure and in-service creep behaviour is not well understood and would be an appropriate topic for a research proposal to the Canadian Electrical Association; CEA are known to be interested in funding work related to thermal plant reliability and life extension.

3. The need for improved ultrasonic inspection techniques to detect Type IV damage has been identified.

6.0 ACKNOWLEDGEMENTS

The author gratefully acknowledges the contributions of the following colleagues to the work presented in this report:

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- Mr. A. Allen for assistance with ultrasonic inspections.
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TABLE I - CHEMICAL ANALYSES

(Courtesy of Chemical Research Dept Reports CE89-1332-K and CE89-3305-K)

Sample	C	Mn	P	S	Si	Cr	Mo	Ni	Cu	As	Sn
3W-C	0.15	0.85	0.018	<0.005	0.42	1.17	0.39	0.087	0.098	0.019	0.006
3W-F	0.10	0.38	0.006	<0.005	0.55	0.97	0.56	0.095	0.10	0.007	<0.005
3E-C	0.15	0.76	0.02	0.02	0.39	1.25	0.53	0.09	0.10	0.02	0.01
3E-F	0.10	0.37	0.005	0.008	0.57	0.95	0.57	0.10	0.11	0.009	0.005
4W-C	0.19	0.59	0.031	0.026	0.28	1.17	0.47	0.13	0.19	0.029	0.017
4W-F	0.10	0.37	0.005	0.008	0.56	1.00	0.57	0.10	0.13	0.01	<0.005
4E-C	0.13	0.79	0.015	0.02	0.32	0.94	0.33	0.13	0.19	0.03	0.02
4E-F	0.091	0.38	0.007	0.01	0.58	1.06	0.57	0.11	0.11	0.011	<0.005
6-C	0.16	0.71	0.023	0.014	0.30	1.19	0.56	0.06	0.063	<0.005	<0.005
6-P	0.11	0.54	0.011	0.005	0.54	0.98	0.58	0.06	0.14	0.019	<0.005
(1) A-335 spec	0.15 max	0.30 0.60	0.03 max	0.03 max	0.5 1.0	1.0 1.5	0.44 0.65	0.40 max			
(2) A-217 spec WC6	0.20 max	0.50 0.80	0.05 max	0.06 max	0.60 max	1.0 1.5	0.45 0.65				
(3) A-182 spec F11	0.10 0.20	0.30 0.80	0.04 max	0.04 max	0.50 1.00	1.00 1.5	0.44 0.65				

(1) piping
(2) casting
(3) forging.

TABLE II

"J" VALUES FOR CASTINGS AND FORGINGS

$(J = (Si + Mn)(P + Sn) \times 10^4)$

Unit 3 West (Failed)	J Value
Casting	304
Forging	102
Unit 3 East	
Casting	345
Forging	94
Unit 4 West	
Casting	418
Forging	<93
Unit 4 East	
Casting	189
Forging	<115
Unit 6	
Casting	<283
Pipe	<173

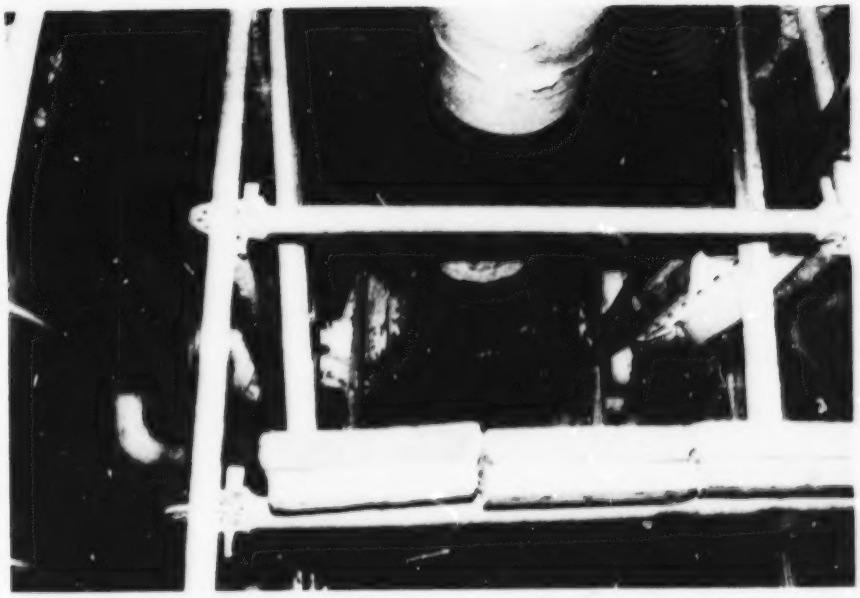


FIGURE 1 - GENERAL LOCATION

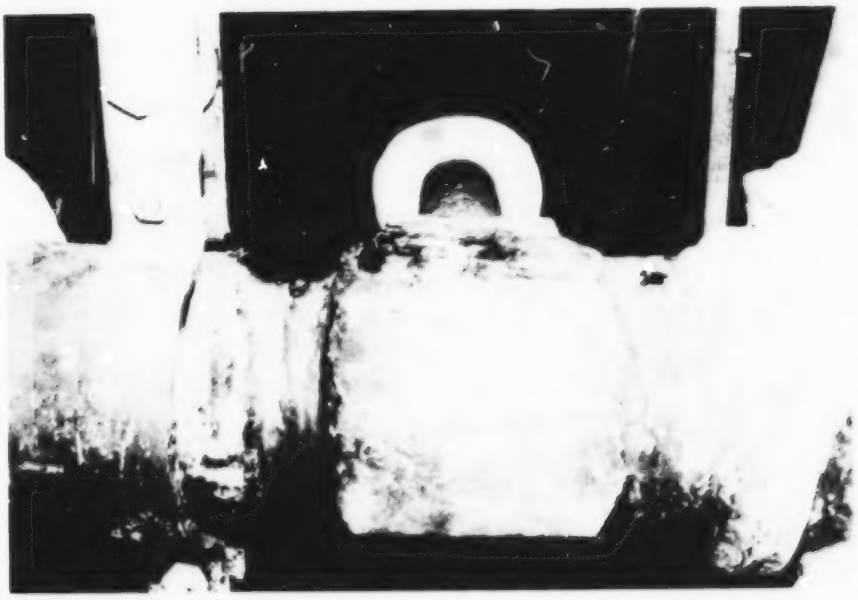


FIGURE 2 - COMPONENT DETAILS:

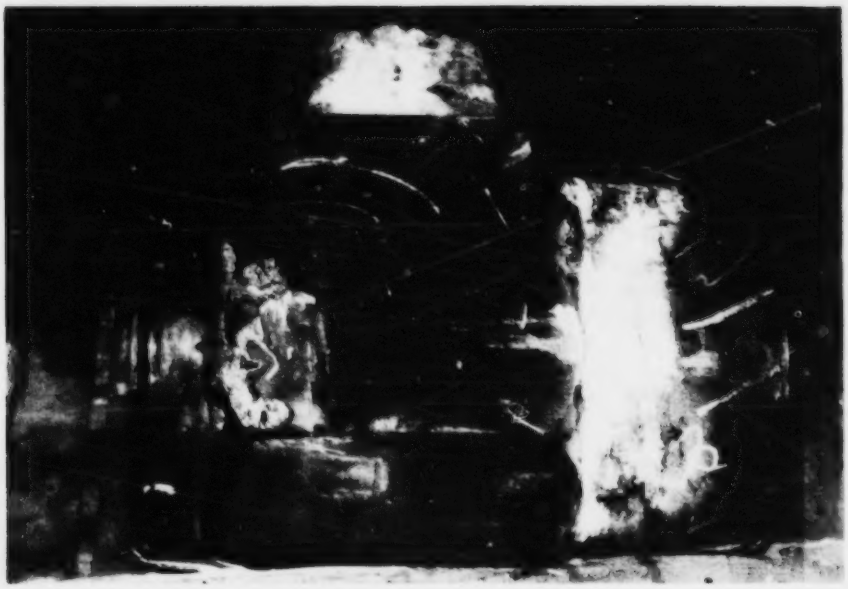


FIGURE 3 - OD CRACK APPEARANCE AT ~6 O'CLOCK POSITION

FIGURE 4 - ID CRACK PENETRATION

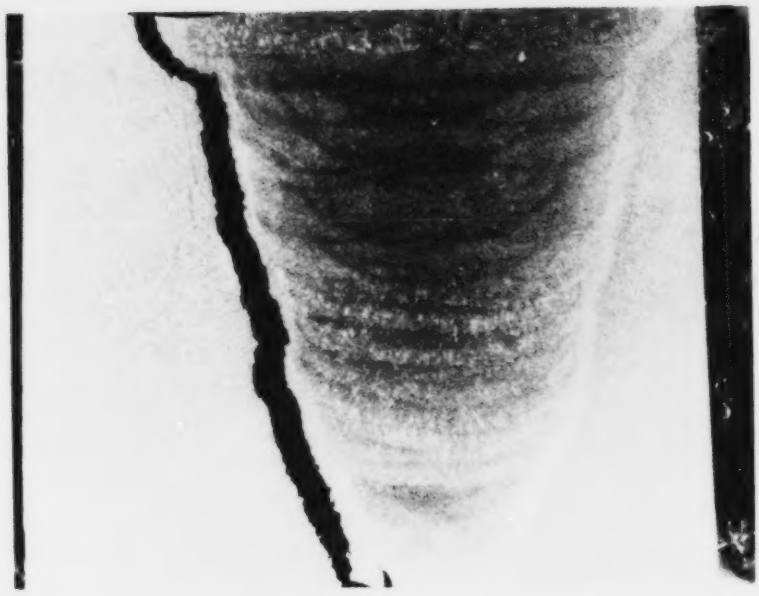


FIGURE 5 - THROUGH-WALL CRACK AT 6 O'CLOCK (X 1.75)

FIGURE 6 - CRACK JUST BREAKING OD SURFACE AT 9 O'CLOCK (X 1.75)

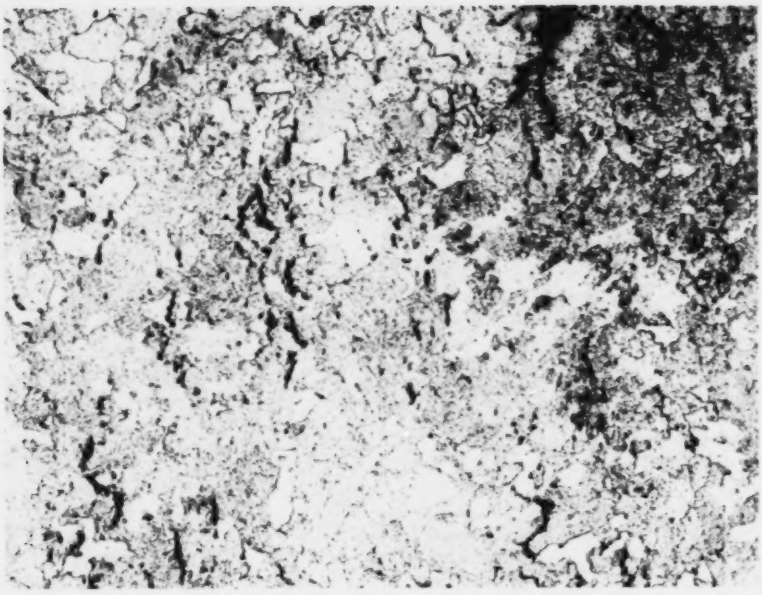


FIGURE 7 - CAVITATION AHEAD OF CRACK TIP AT 9 O'CLOCK (X 400)

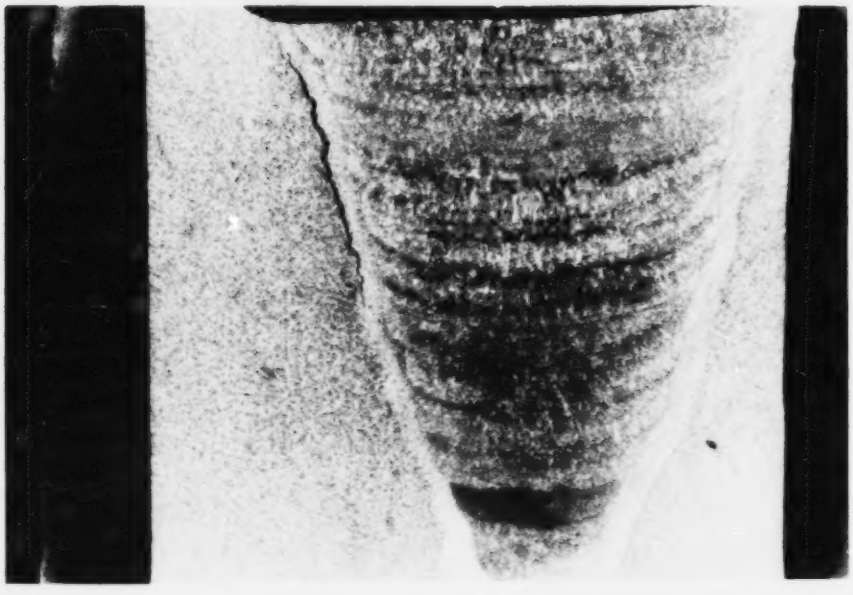


FIGURE 8 - CRACK BELOW OD SURFACE AT ~9-30 (X 1.75)

FIGURE 9 - CRACK TIP REGION NEAR OD AT 9-30 POSITION, X 50

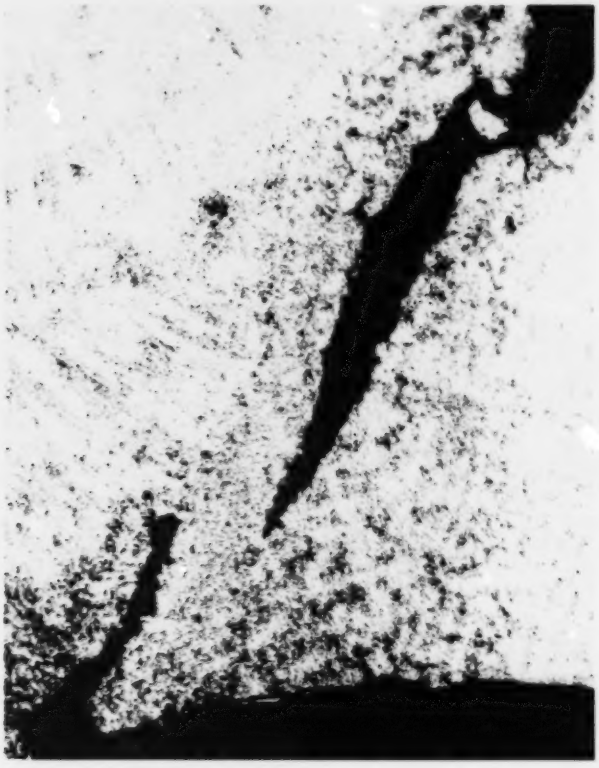


FIGURE 10 - DETAILS OF ID PENETRATION AT 8-15 POSITION, X 12.8

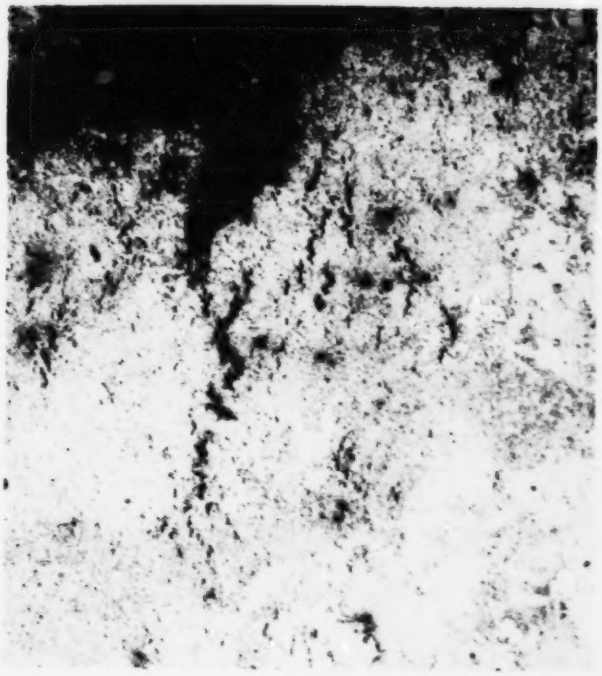


FIGURE 11 - CAVITATION AT TIP OF TYPE IV CRACK, X 128

FIGURE 12 - CAVITATION AT TIP OF TYPE III CRACK, X 128

FIGURE 13 - PROFUSE CAVITATION IN FINE GRAIN ZONE, X 320

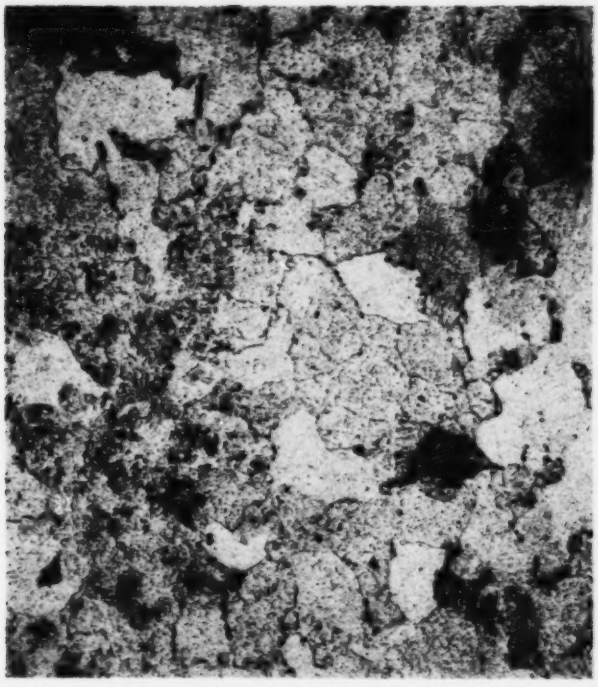
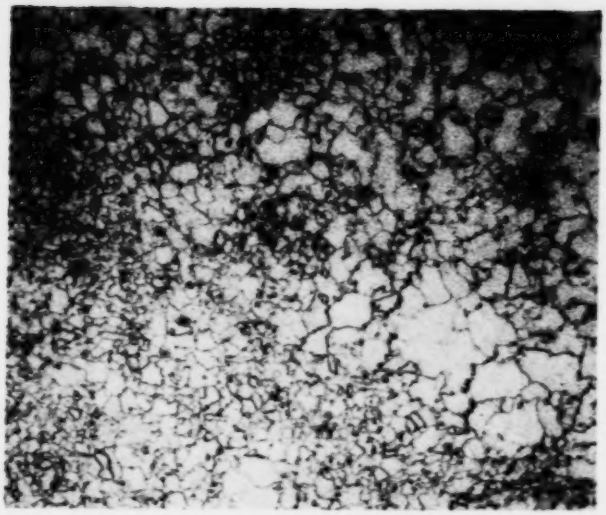
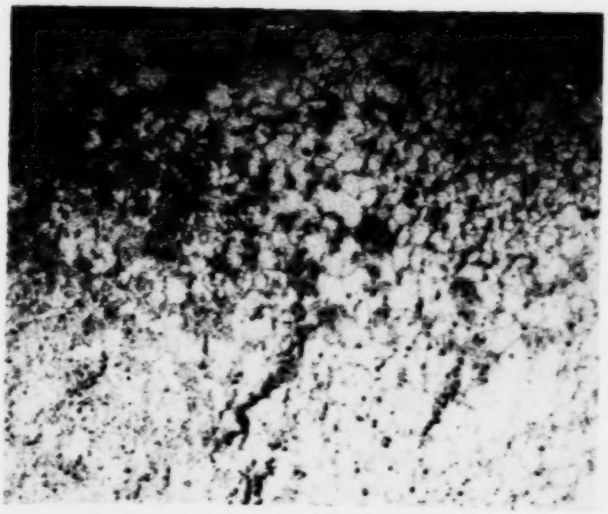


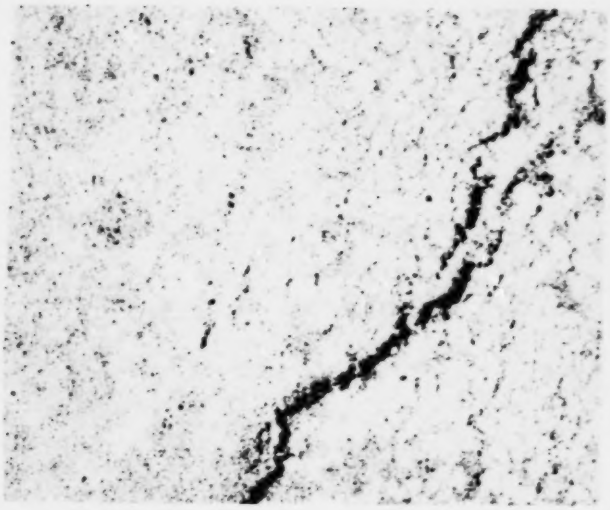
FIGURE 14 - CAVITATION IN ADJACENT PARENT MATERIAL, X 320



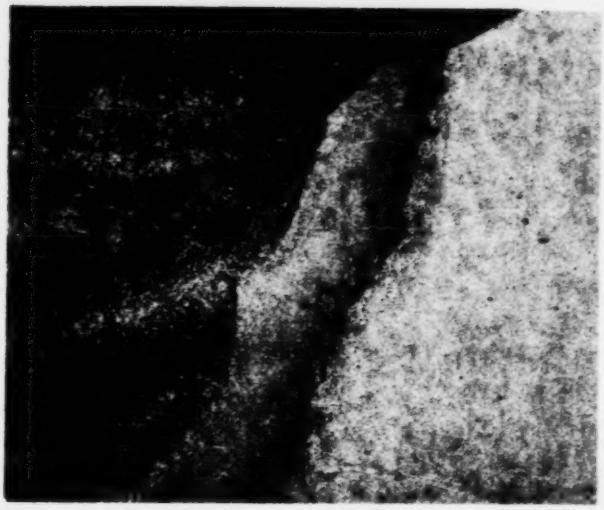
(a) Individual cavities in Type IV region, X 200



(b) Cavities linking to form microcracks, X 100



(c) Higher degree of microcrack development, X 100



(d) Stage preceding final formation of macro-defect, X 7.5

FIGURE 15 - DAMAGE DEVELOPMENT IN UNIT 4 EAST WELD

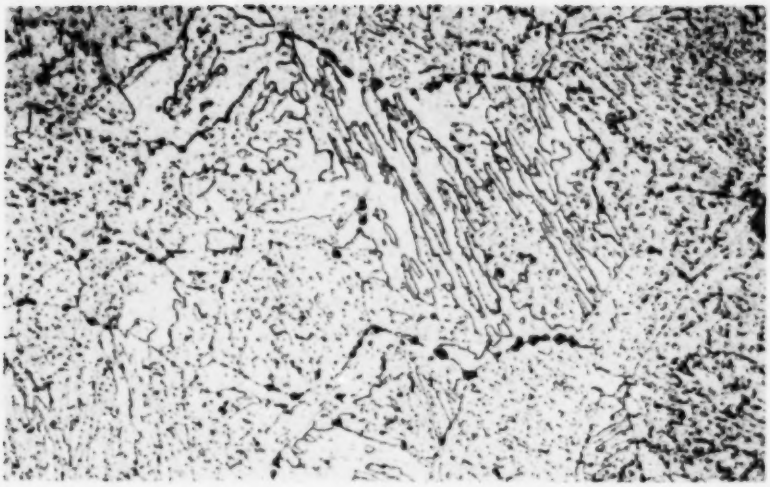


FIGURE 16 - VERY EARLY STAGE IN CAVITATION PROCESS, UNIT 4, WEST WELD, X 800

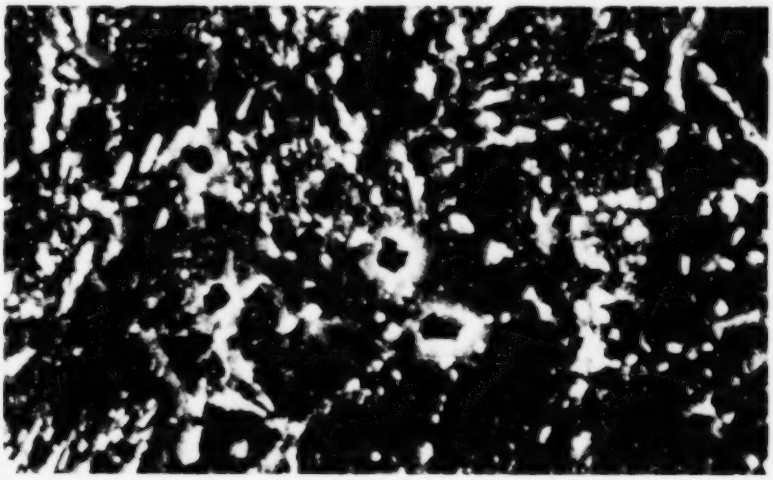


FIGURE 17 - CAVITIES IN SCANNING ELECTRON MICROGRAPH, X 4300

FIGURE 18 - NON-UNIFORM, WAVY NATURE OF CRACK EXTREMITY, X 100

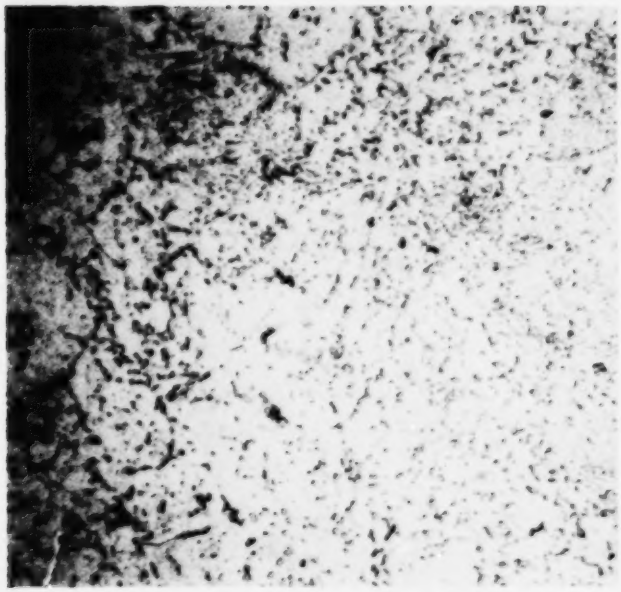


FIGURE 19 - OPTICALLY-RESOLVABLE LIMIT OF CRACK, X 1000

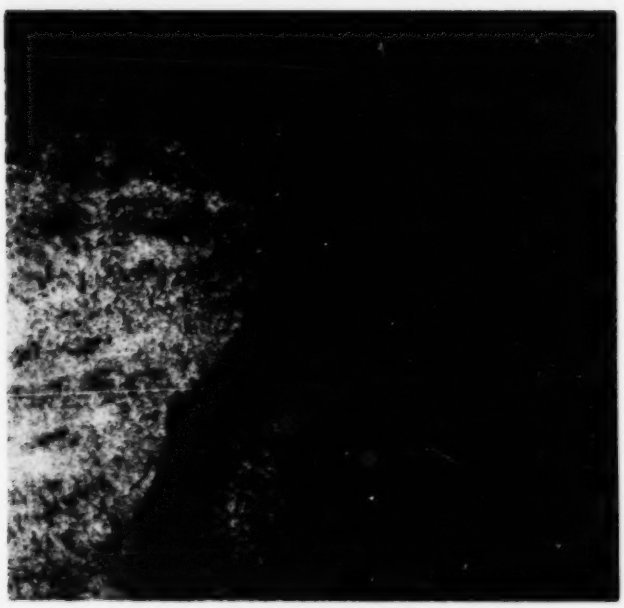
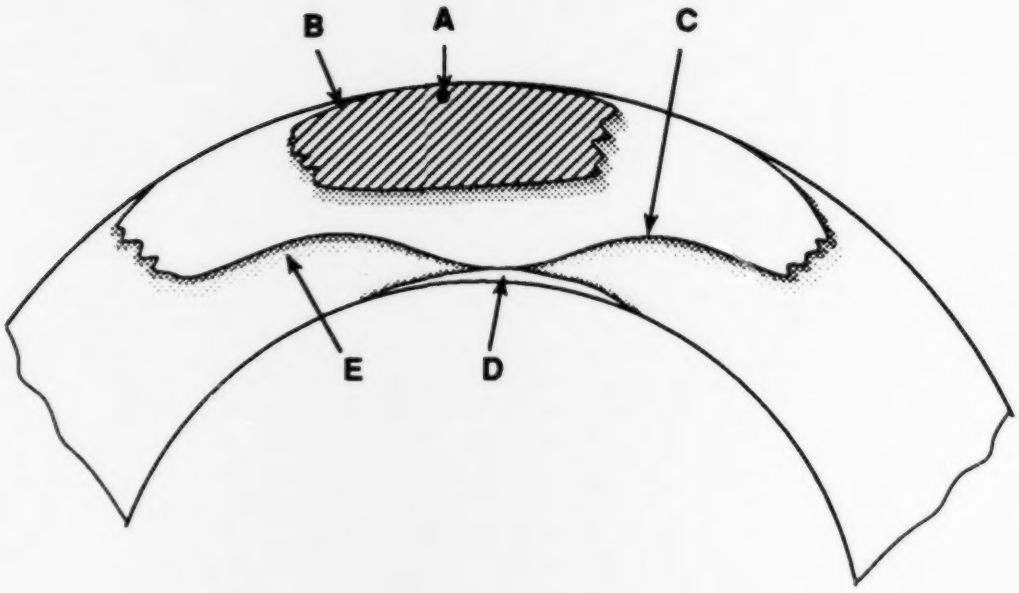


FIGURE 20 - LOCATION OF AREA SHOWN IN FIGURE 19, X 5



A - Defect origin
B - Defect when breaking OD surface
C - Defect at through-wall penetration
D - Crack from ID
E - Cavitation ahead of crack tip

POSSIBLE DEVELOPMENT OF THROUGH-WALL DEFECT
(Based on observations in region of crack extremity)

FIGURE 21

FIGURE 22 - SECTION GIVING LARGE ULTRASONIC INDICATION, UNIT 6 WELD

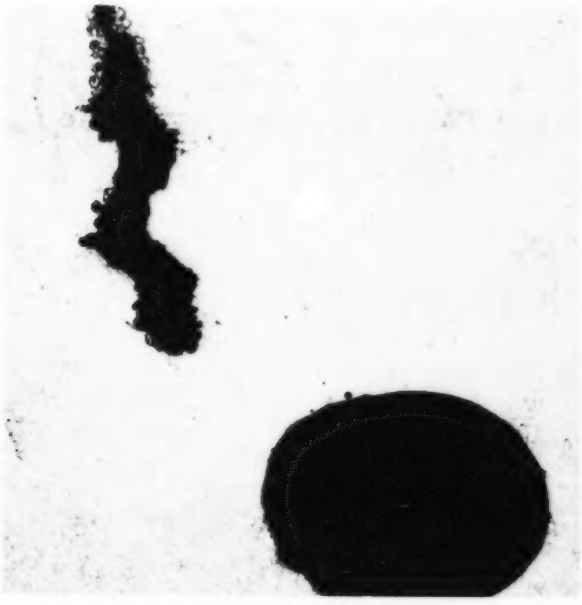


FIGURE 23 - WELD METAL DEFECT RESPONSIBLE FOR U/T INDICATION, X 100

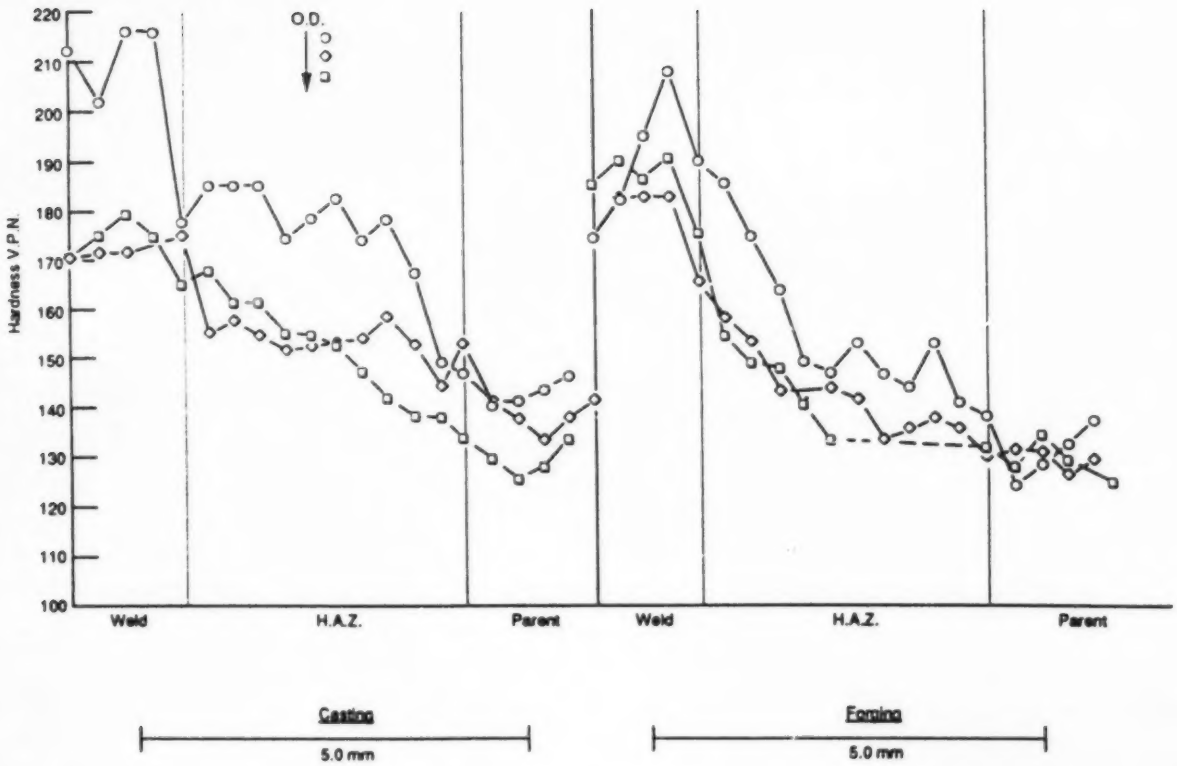


FIGURE 24 - HARDNESS RESULTS ON FAILED WELD AT 12 O'CLOCK (DAMAGE FREE) LOCATION

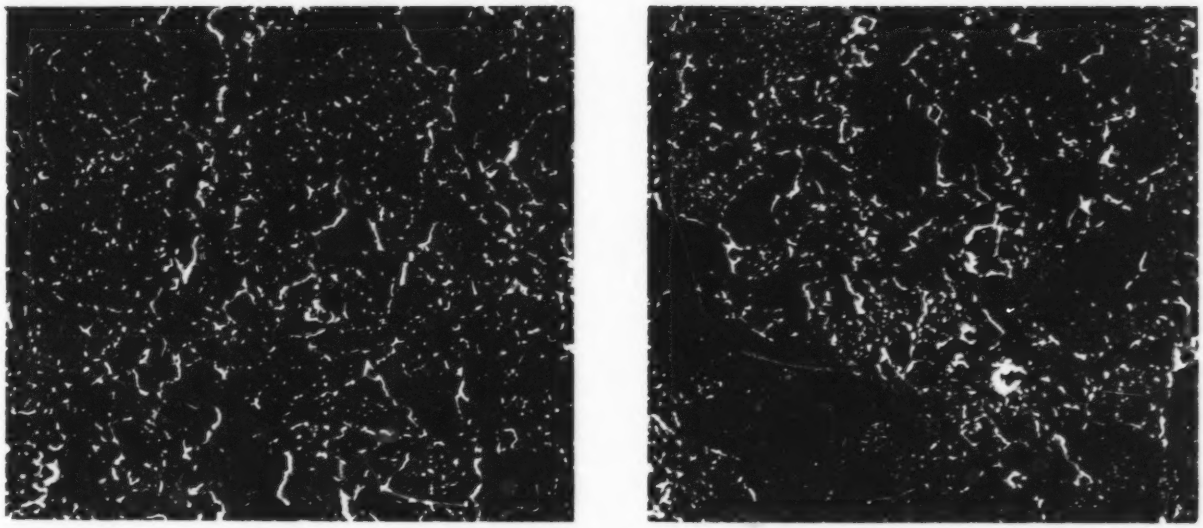


FIGURE 25 - HEAT-AFFECTED ZONE MICROSTRUCTURES IN CASTING, X 100
(a) fine grain zone; (b) partially-transformed zone

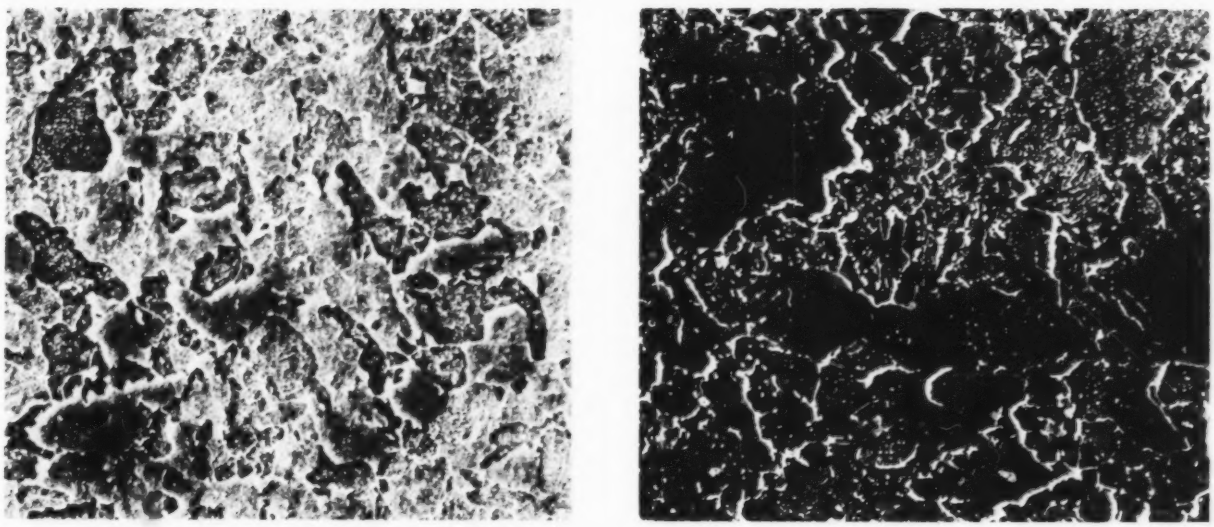


FIGURE 26 - HEAT-AFFECTED ZONE MICROSTRUCTURES IN FORGING, X 1000
(a) fine grain zone; (b) partially-transformed zone

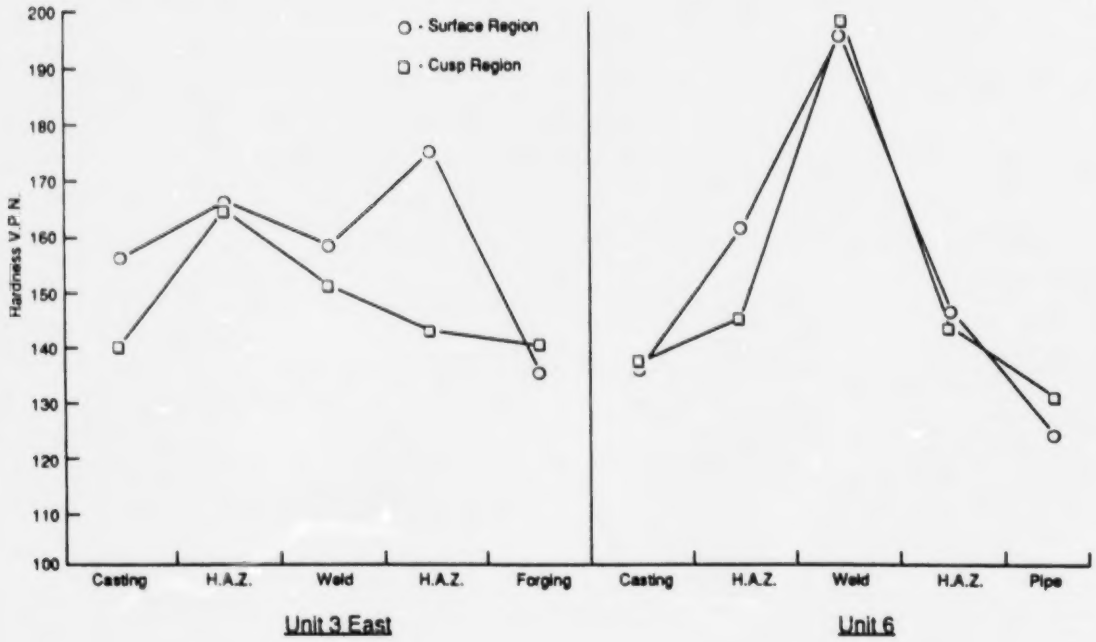


FIGURE 27 - HARDNESS SURVEYS ON DAMAGED (UNIT 3 EAST) AND UNDAMAGED (UNIT 6) WELDS

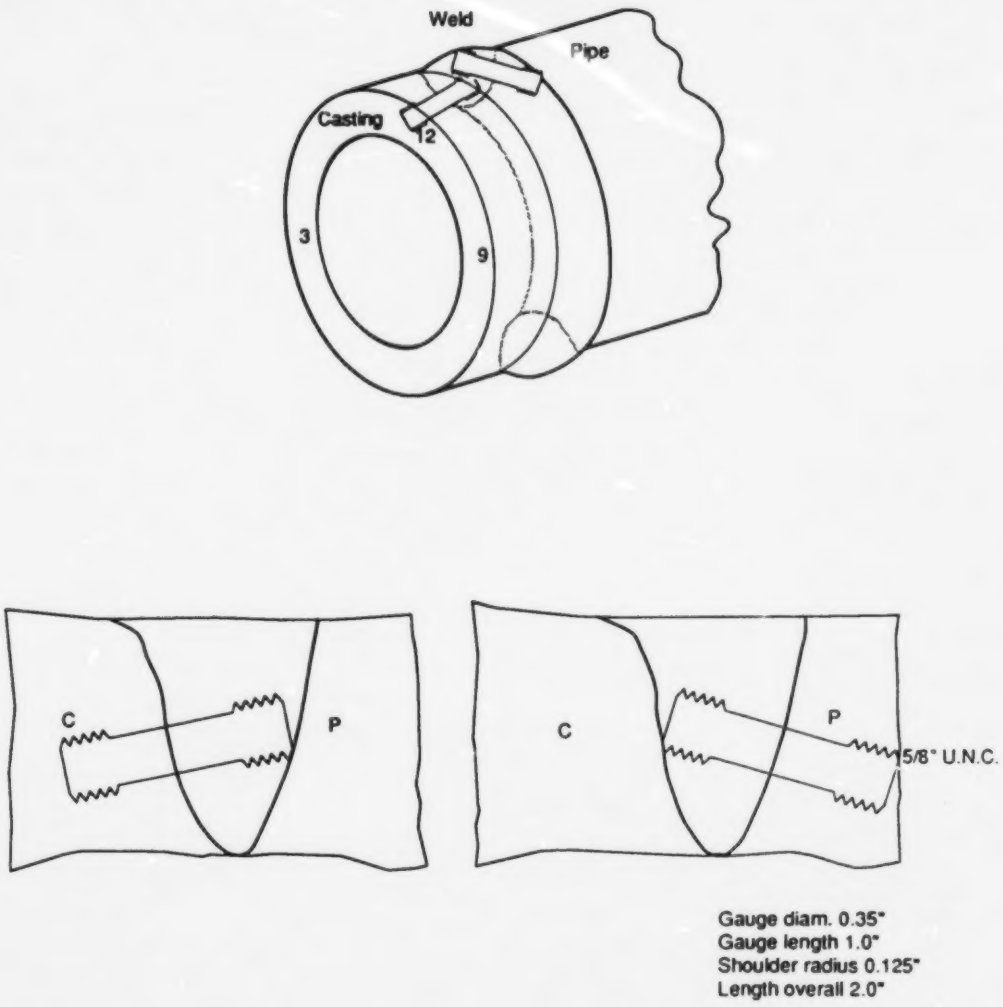


FIGURE 28 - CREEP SPECIMEN DETAILS AND LOCATION

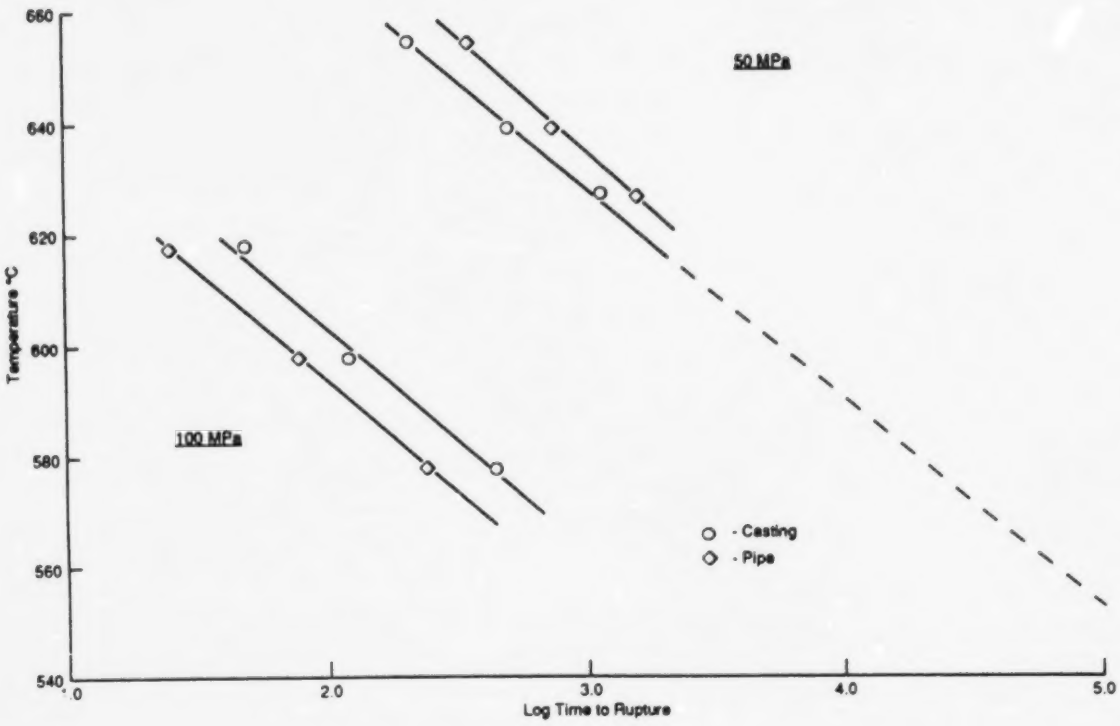


FIGURE 29 - CREEP TEST RESULTS

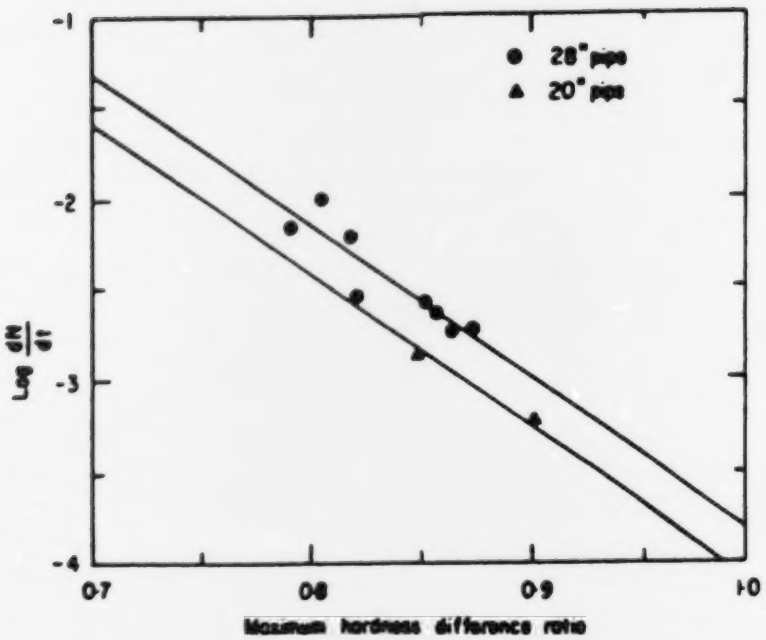


FIGURE 30 - RELATION BETWEEN HARDNESS DIFFERENTIAL AND CAVITATION DENSITY (Ref 7)

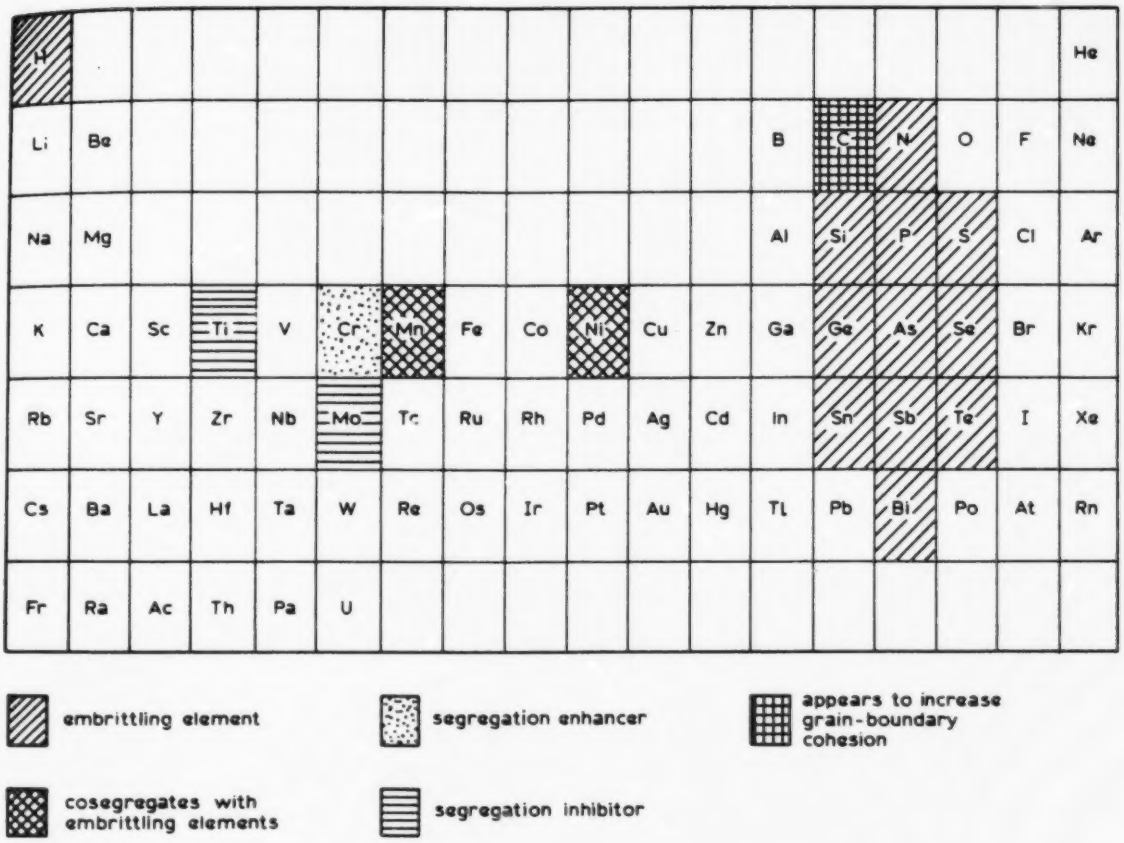


FIGURE 31 - EFFECT OF VARIOUS ELEMENTS ON EMBRITTLEMENT (Ref 11)

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MEMORANDUM

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April 11, 1989

File:823.41(T)

Information Gathering in UK re: Type IV Cracking

The undersigned visited the CEGB Central Electricity Research Laboratories at Leatherhead UK on April 05 for a general discussion of fossil plant life assessment technologies. Opportunity was taken during the visit for a concentrated discussion of their knowledge of the type IV cracking phenomenon in relation to our current experiences at Lakeview GS. To facilitate prompt use of the information gained, this part of the discussion is reported separately in advance. Reference is also made to discussions held on the same subject on April 03 with Dr. Keith Bell of the Engineering Dept of the Welding Institute. Dr. Bell was formerly employed by ERA Technologies Ltd. at Leatherhead on fossil plant materials research.

Present at the specific CERL discussion on type IV besides the undersigned were Drs. Roger Townsend, David Gooch, and Steve Kimmins. Discussions were based on a file prepared by our Dr. H.J. Westwood showing the status of our investigations as of March 24. Due to the nature of the discussions, numerical data below has not been cross checked and should be taken as approximate only.

Type IV cracking has been conceded to be quite an old issue, having been known in the CEGB for more than 10 years. The impression was gained that at least 40-50 joints have been replaced so far due to big flaws or leaks. The CEGB experience has not been characterized by sudden energetic fractures, ie all have been found at a leaking or earlier stage, and none was implicated in personal injury. Bell at TWI provided references indicating the phenomenon has been documented by operators in the USSR and the Netherlands as well as in Germany.

The background R&D effort performed at CERL to provide data for decision making is basically that documented in the 1987 Swansea conference, pp689-703; they do not appear to have gone much beyond this since. There is also some very interesting work (referred to by Bell at TWI) currently underway led by Adrian Williams at Marchwood Engineering Laboratories, on the mechanics of girth joints with a thin hard or soft zone.

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The weakness in creep of the zone centred about the intercritical to just fully transformed HAZ arises not only from the small grain size but from the multiphase structure characteristically found here. The base metals depend for their strength on the (normalizing and tempering) manufacturing sequence that provides a uniform dispersion of intragranular bainitic carbides. Intercritical heat treatment in particular leaves some grains without carbide strengthening. For this reason, every weldment in a ferritic creep-resisting steel must be considered susceptible in principle to type IV damage.

CERL has characterized weldment behaviour by cross-weld creep tests in comparison to tests on base metals alone. The greatest strength loss is associated with the strongest base metals, ie those containing vanadium and/or of high-chromium variety. These materials have high hardenability and hence have a desirable single phase bainitic structure as-manufactured, so they have the "most to lose" from disruption of the carbide structure. For materials like 1/2Cr-1/2Mo-1/4V CERL quotes cross weld creep strengths down to 60% of base metal value. They believe cross weld strength of materials like 1.1/4Cr-1/2Mo will be more like 80% of base metal.

For future plant in the UK, the USA-developed steel T91 (9Cr) is likely to be widely used. This steel is known by CERL to be in the "high susceptibility" group with respect to type IV strength loss in the intercritical zone. No special problems are expected however, in practice, because of the conservative manner in which the material is being applied. The similar German 12 Cr steel "X20 CrMoV 12-1" is in the same category. The biggest practical problem with all the higher Cr creep-resisting ferritics is the lack of a suitable welding filler metal.

Recognizing the characteristic ratio of axial to hoop stress, circumferential welds should still last the life of the pipe in the absence of extraneous axial or bending stress. When they do not, therefore, the immediate concern must always be the source of the excess axial stress. Most of the CEGB's type IV cracking experience is with turbine loop pipes and other comparable relatively short, stiff geometries where substantial axial or bending loads can readily be produced by inappropriate preloading of systems during assembly. Efforts have been made to estimate loadings on several damaged assemblies by measuring springback during loop disassembly, and they state it as a general finding that where type IV damage is found, the cause is usually clearly evident during this process.

For the above reasons, the CERL staff were concerned at the Lakeview situation, in that it does not appear to fit the established pattern. That is, the timing of 80,000 hours seems early for a nonproblematic vanadium-free steel, and the failures are occurring at the end of a long and relatively flexible loop. They pointed immediately at two features of our situation as questionable. First, the stop valve castings all have a circumferential ring of extra material adjacent to the weld preparation. The strain concentration arising from the change of section at the edge of this reinforcement is in about the worst possible location. Bell at TWI mentioned the same concern and specifically suggested that this reinforcement be machined away insofar as possible. Secondly, it was felt highly likely that some unanticipated axial or bending force was acting at temperature due to poor design or poor presetting of the hanger system in the vicinity. The broken hanger on the Unit 3 west pipe may not be the cause of the failure but it is surely a sign of large thermal movements for which inadequate provision had been made.

If it should transpire that there is nothing very remarkable about the axial or bending loads in the failure area, then that is also a concern in that it would signal the fact that these pipes (including base metals) are near the end of their useful life.

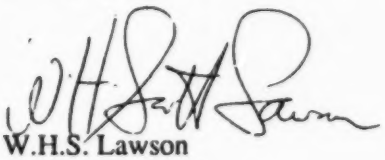
Discussion continued on the propensity of type IV damage to nucleate and grow buried within the pipe wall and the implications of this for decision making when type IV is found at the surface. Theoretical justification for the buried growth characteristic in a situation free of bending, has been documented by Williams and by workers in the USSR. Papers provided by TWI are available on request to the undersigned. CERL seem merely to have accepted this as a given, and seemed not to have data as to how big the stress gradient due to bending has to become, to move nucleation out to the surface. This of course forms the basis for the general CEGB procedure that when type IV damage is found at the surface, the weldment is to be assumed to be in an unacceptable condition unless there is other definitive evidence of extreme bending or other clear sign of a strictly superficial phenomenon. Usually they find that a HAZ with surface breaking damage is sufficiently badly cavitated internally that appearance of a macroscopic defect is likely prior to the next scheduled periodic inspection.

Recognizing the weakness of surface examination for early detection of type IV damage, volumetric NDE is of major importance to the CEGB. Ultrasonic examination is the normal inspection method, and all modern weldments have an appropriate surface finish and profile for U/T. Older stations were generally subjected in the 1970's to intensive U/T examination for coarse grain HAZ (reheat type) cracking and the weldment surfaces were usually reprofiled at that time, so that minimal new effort is now required. Also related to periodic NDE, CERL staff stated that the formal periodic inspection procedure for high energy piping includes examination of the integrity and preloading of hanger systems as well as the usual flaw search procedures on the pipes themselves.

A major opportunity exists for R&D on volumetric NDE methods specific to the profuse fine cavitiation (typically less than 10 microns diameter) that characterizes the mid-life state of type IV damage prior to the emergence of a macroscopic flaw. The CERL staff cited work ongoing in CEGB on U/T velocity changes and on measurements such as grass level. But one gets the impresssion they have nothing nearly ready for field use.
The undersigned attempted to get an impression of whether composition/minor element/ impurity effects are known to be related to the likelihood of type IV damage. The focus was on reasons why all our problems had shown up in cast material. They repeatedly opined that it is probably not the chemistry but is likely the mechanical geometry that is making the casting side HAZ appear weaker than the mating HAZ in forging or pipe.

Bell at TWI was of the opinion that durability problems have been experienced with repairs of type IV damage. As evidence, he cited a Dutch paper in the October 1987 Inst of Mech Eng Conference "Refurbishment and Life Extension of Steam Plant" . Those authors cite a repair lifetime of 30-40% of the original weldment life (copy of paper available from the undersigned). The CERL staff were questioned closely on this issue.

Their comments are that certainly one can expect problems if the material in which the repair HAZ is placed was already cavitated. It is always a good idea to machine back a couple of inches from the original weld location if possible to minimize this problem. They also point out that most type IV repairs are done to base metals that have already been in service for more than 60% of their design lifetime. They see no reason based on HAZ microstructure why the intercritical HAZ on a used material ought to be grossly weaker than that on a new material, and this is borne out in practice in that they are not seeing repeat failures of repairs provided the original axial stress problem was solved during the repair work.



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